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STUDIES OF HEAT-RESISTANT MATERIALS FOR HIGH-TEMPERATURE GAS REACTOR

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16. Abstract As one of the countermeasures against the energy crisis, hope is entertained for the development of nuclear energy. MHD power generation may be of assistance from the standpoint of improved energy efficiency; high-temperature gas reactors are noted for their high efficiency. The GGA Company has developed a large-scale commercial high-temperature gas reactor whose prospects for the future are good. A preliminary requirement for using high-temperature gas economically and safely is the development of heat-resistant materials. In this report, data are summarized concerning the suitability for this purpose of the existing alloys, including the alloys which are currently being mentioned as candidates. In tables and figures, data on corrosion resistance, creep resistance, thermal fatigue resistance, structural stability, weldability and machinability are reported.

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STUDIES OF HEAT-RESISTANT MATERIALS FOR HIGH-TEMPERATURE GAS REACTOR

Shōzō Sekino

I. Introduction /1*

The energy crisis is today creating an immense uproar all over the world, particularly in the United States. One aspect of this is the oil problem. A second aspect is the question of the total amount used; any greater density of energy use in the excessively congested industrial areas would greatly disturb nature. As for the countermeasures against these problems, hopes are being entertained for the development of nuclear energy as a means of solving the oil problem. With respect to the second aspect, there is hope that MHD power generation may be of assistance from the standpoint of improved energy efficiency. In the field of nuclear power generation itself, hopes are attached to high-temperature gas reactors, which are noted for their high efficiency, and it is also hoped that it will be possible to make direct use of nuclear energy. The GGA Company has already developed a large-scale commercial high-temperature gas reactor, and the future prospects are bright. If one were to make use of thermal energy directly in various processes without passing through the medium of electricity, it would be desirable to have a gas temperature of at least 1000°C. In the project now being planned by the Ministry of International Trade and Industry for multipurpose utilization of nuclear reactors and for direct steelmaking, the target is also 1000°C. A preliminary requirement for using such high-temperature gas economically and safely is the development of heat-resistant materials. In this report we have summarized chiefly our own data concerning the suitability for this purpose of the existing alloys, including the alloys which are currently

*Numbers in the margin indicate pagination in the foreign text.

being mentioned as candidates.

II. Material Characteristics Required for High-Temperature Gas Reactor Use

According to the current plans, helium gas is the coolant for high-temperature gas reactors, and it is supposed that there will be no change in this in the future. At 1000°C, helium gas assumes the form of a reduction gas, for example as He → He or He → H_2 + CO, etc., or it may undergo heat exchange and be transformed into H_2O. These gases are used in the chemical industry, in the steelmaking industry, or in power generation. Consequently, the heat-resistant materials are required to have the following characteristics in these atmospheres, including also the welds:

i) corrosion resistance,
ii) creep resistance,
iii) thermal fatigue resistance,
iv) structural stability.

At the same time, with respect to the manufacturing considerations, they are also required to have:

v) weldability,
vi) machinability.

If a material is deficient in even one of these characteristics, the material cannot possibly be used. These characteristics must also have a high reliability. Once an accident has occurred in a reactor, the social consequences would be extremely far-reaching.

III. Corrosion in He and CO + H_2 at 1000°C

In Table 1 are shown the chemical compositions of representative superalloys universally used as high-temperature materials [1]. In Fig. 1 are shown the results of corrosion tests of these alloys performed at 1000°C in helium with a purity of 99.99%, in air, and in 65% CO + 35% H_2. Strangely enough, the heat-resistant steels tend to corrode easily in helium, and there are some alloys, such as, for example, Incoloy 807, in which this tendency is /2 pronounced. Naturally, the alloys are not corroded by helium, but rather by the O_2, which is contained at a rate of 0.01% or less in the helium. This tendency resembles the results for corrosion in a low vacuum; if the helium purity is increased in the testing, the alloys will not be corroded at all. In actual fact, it is believed that the amount of gaseous impurities in a high-temperature gas reactor will be less than 30 ppm. However, it is true that the oxidation potential of a gas fluctuates greatly depending on the gas composition, and if this fact is taken into consideration, it is difficult to estimate the degree of corrosion in an actual reactor. Figure 2 shows the changes in the Cr quantities from the boundary plane with the oxidized layer toward the matrix [2]. It is clear that dechromization tends to occur readily in helium and that the oxidized layer lacks fineness and is unable to protect the matrix. The results of EPMA analysis of the oxidized layer are shown in Fig. 3 [2]. In view of the oxidation potential of the atmosphere, Fe is not oxidized in helium, and the oxidized layer consists chiefly of Cr_2O_3 and is porous. Under such conditions, it may be considered natural that those elements which are easily oxidized, such as Al and Ti, should be intensively and selectively oxidized. The degree of dechromization in various types of superalloys is shown in Fig. 4. There are big differences depending on the system of components [2]. An examination of the relationship with the components indicates that the depth D of the dechromization region is indicated by

$$D = 14.8 + 4.72Cr + 52.5Ti$$

and it is clear that it is governed by the Cr and Ti contents. The relationships between the calculated values and the measured values are shown in Fig. 5.

The fact that the superalloys have a poor corrosion resistance in impure helium is unfortunate with respect to high-temperature gas reactors. Corrosion was allowed to take place experimentally in the air to form a fine oxide layer on the surface. Then studies were made to determine whether this layer was able to play the role of a protective layer in helium. The results are shown in Figs. 6 and 7. An effect is clearly observable in certain alloys. However, when one remembers that oxide layers formed in air cause changes in the composition in a helium atmosphere, it is doubtful whether this effect can continue for a prolonged period of time. In Fig. 8 are shown the results obtained when an even more thorough-going approach was adopted and a heat-resistant coating was applied on the surface. The calorizing effect is extremely pronounced. However, many more studies will be necessary before it becomes possible to put it into actual application. For instance, it will be necessary to study whether it can be used stably over a long period of time and also to study the spalling resistance and other matters.

IV. Creep Strength

The estimated rupture strengths of various alloys at 1000°C and 10^3 hours are shown in Table 1. In the National Project, a strength of more than 1.0 kg/mm^2 at 1000°C and 10^5 hours is considered for the superalloys. The superalloys meeting these standards are Supertherm, Nimonic 115, Udimet 700, René 41, TAZ8A, and Astroloy. However, since all of these alloys, with the sole exception of Supertherm, are γ'-reinforced alloys, the long-term stability of the γ' is of importance. Furthermore, as will be mentioned later on, γ'-reinforced alloys have a poor weldability. At 1000°C, γ' is already unstable and has rapid growth. In Fig. 9 are shown the stress-rupture relationships for Ni-base alloys which are solid-solution-reinforced alloys (Hastelloy X, C), solid-solution + precipitation-reinforced alloys (Inconel 617, René 41, Astroloy), and dispersion-reinforced alloys (In. 853, TDNi). In view of this, one may say that solid- /3 solution-reinforced alloys have a relatively stable creep strength for a longer time than γ'-reinforced alloys. Dispersion-reinforced type alloys display an extremely high, stable rupture strength. However, besides the problems in manufacturing and welding, economically their costs are too expensive, and it will require a considerable time before they can be put into actual application.

The experiments described above are the results in the atmosphere. However, is the creep strength in helium the same as the values in the atmosphere? In Fig. 10 are shown, by way of example, the rupture times at 1000°C of Inconel 617. In helium containing a certain amount of O_2, the rupture times may decrease to one-third of those in the atmosphere. If the purity of the helium is increased, the rupture time returns to the value in the atmosphere [3]. If these facts are considered in connection with the previous data on corrosion, it appears best to interpret them as follows. That is, the fact that the rupture strength decreases in impure helium may be attributed to the intense internal oxidation, which forms the point of origin of a crack and causes decreases in the strength. At any rate, it is believed that more thorough-going studies will be necessary in the future concerning the effects of the compositions of the gases present in minute quantities and also about creep under variable stresses.

V. Thermal Fatigue

In an atomic reactor, one must assume that there will be at least several shutdowns a year on account of accidents or for periodic inspections. Consequently, an estimate that thermal fatigue will occur 500 times in one lifetime may be regarded as quite safe, or perhaps even erring on the safe side. Figure 11 shows the stress applied until rupture occurred 500 times for various superalloys which were subjected to cycles (each cycle lasting a total of 8 minutes) consisting of heating between 200-1000°C, holding, cooling, and holding, each for 2 minutes, while a constant stress was being applied. It is noteworthy that the high C alloys such as supertherm or HK40 have a poor resistance to thermal fatigue. Figure 12 indicates that the thermal fatigue strength can be explained more or less in terms of the hot elongation and the rupture strength at 1000°C for 1000 hours.

Even with the same components, quite different results for the thermal fatigue are obtained when different melting practices are used. The results for Incoloy 800 are shown in Fig. 13. The alloy obtained by the VIM-ESR method showed the best characteristics [4]. This also can be explained in terms of the above-mentioned reasons.

When the thermal fatigue in a vacuum is compared with that in the atmosphere, it is well-known that there is a much longer lifetime in the former case [5]. As the reason for this, it is mentioned that rewelding is difficult because the fresh surface is oxidized. The question is this: to what degree is the thermal fatigue affected in impure helium? According to the results of Kondo et al. [6], fatigue cracks are propagated more rapidly in helium than in the atmosphere, and the breaking mode also changes from the through-grain type to the grain boundary type. According to the corrosion data, there is intense internal oxidation along the grain boundaries in impure helium. It is thought that the grain boundaries which have undergone this internal oxidation become the points of origin of the cracks and also assist the propagation of the cracks. From this standpoint, it may perhaps be possible to say that alloys which do not easily undergo internal oxidation are also desirable alloys with respect to the heat fatigue resistance. According to Wells et al., the fatigue life is improved considerably by aluminized coating, and one may also have to think of applying surface treatment if this becomes necessary in the future.

VI. Weldability /4

Superalloys are difficult to weld. The higher the strength of a material, generally speaking, the easier it will be for hot cracking to occur. Even if welding is possible, the creep strength of the weld part may not equal that of the parent metal, or even if there is sufficient strength, the creep elongation may be extremely small. It is believed that weld cracking is caused by the formation of compounds with a low melting point on account of a eutectic reaction on the grain boundaries. In Table 2 are shown, by way of examples, the eutectic points of mainly Ni compounds. Figure 14 shows the influence of Zr on the hot cracking of Inconel 600. When Zr is added, hot cracking becomes pronounced. Although Zr is a desirable element for improving the hot workability and the creep resistance, it is difficult to put it into actual use because it impairs the weldability.

The next problem is that of strain-aging cracking. In γ'-reinforced alloys, when welded metals are reheated, precipitation of γ' will occur in a definite temperature region, and cracking will occur on account of the stress at this time. The creep strength can be heightened by increasing the amounts of γ' precipitated. However, precipitation of γ' commences already during the cooling after welding, and a sort of strain-aging cracking occurs. Figure 15 is an example of this, indicating that when the Al and Ti quantities in Astroloy alloys are increased, weld cracking tends to occur remarkably easily.

Among the welding methods applied, there are manual welding, MIG welding, TIG welding, and EB welding. The last-mentioned methods are the more desirable methods, in which there is also a high purity of the weld metals. In particular in EB welding, the drawback that creep rupture elongation of the weld metal is difficult to obtain is eliminated to a considerable degree. However, in each of these methods, the alloys are melted and solidified, and the drawbacks inherent in solidified structures have not been eliminated. A method called Transient Liquid Phase Bonding was recently announced by the Pratt & Whitney Company. It is believed that much can be anticipated in the future from such diffusion-bonding methods.

VII. Hot Workability

The higher the strength of an alloy, the more difficult it is to work. This is even more difficult in metals for high-temperature gas reactors, in which long pipes with narrow diameters are required. All cracks during hot working of alloys are on the grain boundaries, and cracking is determined by the relative strengths of the matrix and the grain boundary and by the brittleness of the grain boundary. Fundamentally, it is determined by the strength of the alloy, but since the grain boundary has a relationship to this, the influence of the microelements is extremely great. By way of an example, the influence on Inconel 600 is shown in Fig. 16. Zr and Ce are precipitated onto the grain boundaries, and it is thought that they cause morphological changes in the sulfides and the oxides which are present here, strengthening the grain boundaries [8]. Figure 17 shows the distribution of B according to the fission track etching method with respect to Inconel 600 to which B was added. It is proved that B is precipitated preferentially and forms two layers on the grain boundary.

The workability also varies greatly depending upon the melting practices. In Fig. 18 are shown the results obtained by

Pridgeon et al. [9], who compared the reduction of sectional area at high temperature of Udimet 700 which had been melted by the ESR and VAR methods. It is clearly indicated that the ESR melting method is superior. Figure 19 shows the differences in the hot workability of Hastelloy X when different melting practices were applied. It is clear that the workability increases remarkably with the ESR melting method. In ESR melting, the amounts of S and O become extremely low, and it is believed that this causes an increase in the melting point of the /5 precipitates on the grain boundaries and strengthens the grain boundaries, thus resulting in an improvement of the hot workability. It is also true that the hot workability differs considerably with different slag systems. There are many matters in this regard which will require study in the future.

Table 3 shows the results obtained by R. Schlatter [10] when he studied the optimum melting practices for superalloys. He holds that the AM-AOD process and the plasma-arc melting method are the methods with the greatest future promise as the primary melting method and the secondary melting method, respectively.

No matter what method may be adopted, the hot workability of superalloys will absolutely continue to be as bad as before. Therefore, it will be necessary to proceed also with studies of the working methods themselves, for example, of the hydrostatic extruding method and the integral ingot-making method.

VIII. Comprehensive Discussion and Future Prospects

We have gathered together some of the data for alloys which are being considered currently for application to high-temperature gas reactors and for super-strength alloys which have been publicly announced for forging purposes. We have investigated their suitability for use in high-temperature gas reactors. The results are shown in Table 4. If an alloy here has an insufficient creep strength, its economic properties will drop, but this can be compensated by increasing the pipe thickness. It is highly possible that the corrosion resistance and thermal fatigue resistance in helium can be increased by surface treatment. In the final analysis, at the present stage the characteristics which ought to be viewed with the greatest importance are the weldability and the hot workability, since nothing at all can be done about the inherent characteristics of the alloys themselves. Also shown are the results of comprehensive evaluations, taking these points into consideration. Inconel 617, Incoloy 800, Inconel 600, and Hastelloy-X may be regarded as the most promising among the alloys. It is possible to standardize more or less the fundamental component systems of these alloys and to heighten their hot workability or their corrosion resistance by adding microelements. Therefore, in the future, it will be necessary to move ahead with studies along this line.

With respect to the corrosion resistance, it is probable that surface treatment will lead to the final solution. In this connection, Co-Cr-Al-Y surface coatings are today being applied to jet engine blades, and it is encouraging to note that they have withstood up to 20,000 hours or more.

An even more important characteristic, which was not mentioned in the explanations given thus far, is the long-term structural stability. In particular, when held under stress for prolonged periods, TCP phases sometimes occur and lead to pronounced embrittlement. This point will also have to be studied thoroughly.

With respect to welding, EB welding is the most promising method at the present time, and it is believed necessary to clarify the EB welding conditions and the characteristics of the EB weld which will not produce cracking.

As for the production properties, it is by all means necessary to clarify first of all the melting conditions which will increase the hot workability, and also to develop working methods by which it will be possible to fabricate even those materials with a poor workability.

It may be said that if all of these studies are carried forward concurrently, it is highly possible that the goals of the National Project can be achieved.

TABLE 1. CHEMICAL COMPOSITIONS AND HIGH-TEMPERATURE CREEP RUPTURE STRENGTHS OF REPRESENTATIVE HEAT-RESISTANT ALLOYS (TANAKA RYOHEI).

Typical super alloys and their properties

表 1. 代表的な耐熱合金の化学成分と高温クリープ破断強度（田中良平氏）

	Alloys	Chemical Composition												Creep Rupture Strength (kg/mm²)					
														900℃		1,000℃		1,100℃	
		C	Cr	Ni	Co	Mo	W	Ti	Al	Si	Mn	Fe	Others	30,000 hr	100,000 hr	30,000 hr	100,000 hr	30,000 hr	100,000 h
Fe-Base Alloys	Incoloy 800	0.05	20	32				0.5	0.5					1.1	0.9	0.6	0.4		
Fe	LCN 155	0.15	21	20	20	3	2.5						1Nb, 0.15 N 0.1 Cu	2.5	1.5	0.8	0.5		
	Incoloy 901	0.05	14	43		6.2		2.5	0.25					2.5	1.5	0.8	0.5		
	HK 40 (C)	0.4	25	20										2.0	1.6	1.2	0.8	0.4	0.3
	Supertherm(C)	0.5	26	35	15		5							4.5	3	1.7	1.1	0.6	0.4
Ni-Base Alloys	Inconel 600	0.04	16	残								7.2	0.1 Cu	1.0	0.8	0.5	0.3		
	Inconel X	0.04	15	〃				2.5	0.9			2	0.9 Nb	2	1.5	0.2	0.1		
	Hastelloy X	0.1	22	〃	1.5	9	16					18.5		2.4	1.9	1.1	0.8	0.5	0.4
Ni	Nimonic 80 A	0.1	20	〃	2.0			2.2	4.5			1		0.8	0.5				
	〃 90	0.1	20	〃	18			2.5	1.5			5		2	1.5	0.2	0.1		
	〃 100	0.3	11	〃	20	5		1.2	4.5			1		3.5	1.7	0.4	0.2		
	〃 115	0.2	15	〃	15	3.5		4	5					6.5	4.5	2.8	2	1	0.7
	Udimet 700	0.15	15	〃	18.5	5.2		3.5	4			1		6.5	4	1.8	1	0.5	0.3
	René 41	0.1	19	〃	11	10		3.1	1.5				0.01 B	5	4	1.8*	1.5*	1	0.8
	〃 62	0.05	15	〃		9		2.5	1.3			22	2Nb, 0.01B	2.5	1.5	0.8	0.5	0.3	0.2
	TAZ 8	0.12	6	〃		4			6				8Ta, 1Zr, 2.5V	13	9	5	3	1.3	1.0
	Astoroloy	0.06	15	〃	15	5		3.5	4.4				0.03B	11	9	3*	2*	3	2.5

*Estimated from data of the writers.

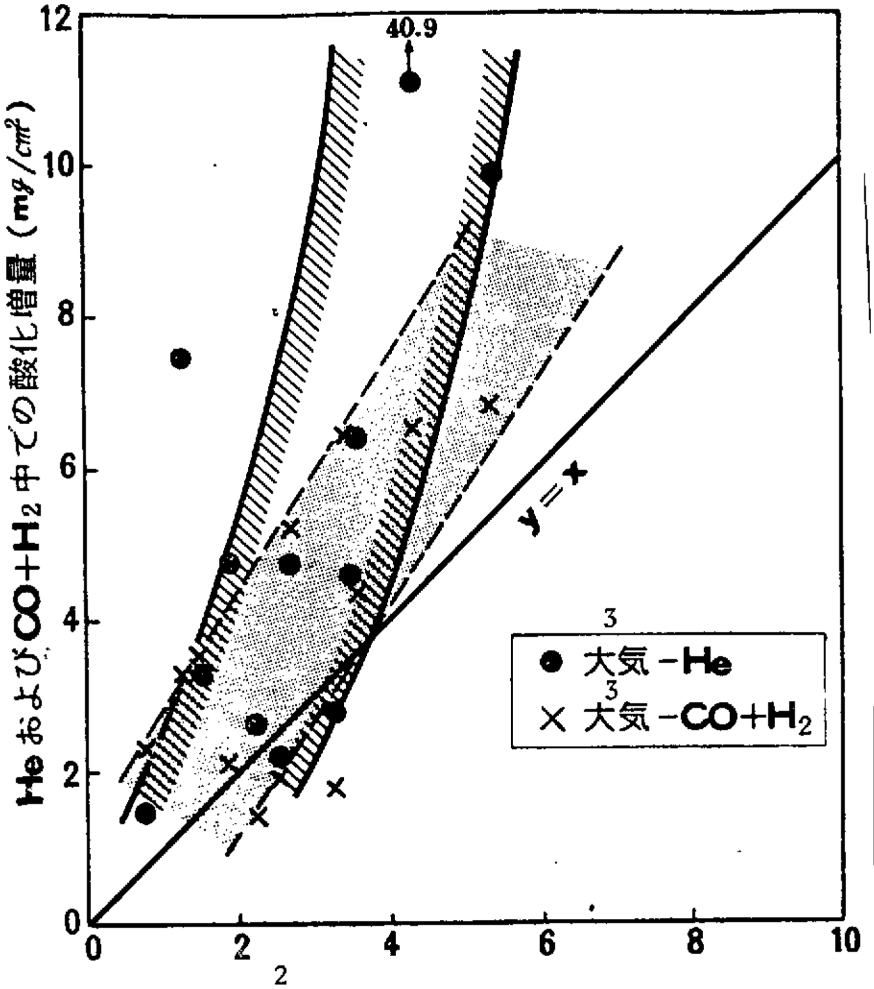

Fig. 1. Comparison of corrosion data of various super-alloys in the atmosphere and in He and the atmosphere [sic].

Key: 1. Weight increase by oxidation in He and in CO + H_2
2. Weight increase by oxidation in air
3. Atmosphere

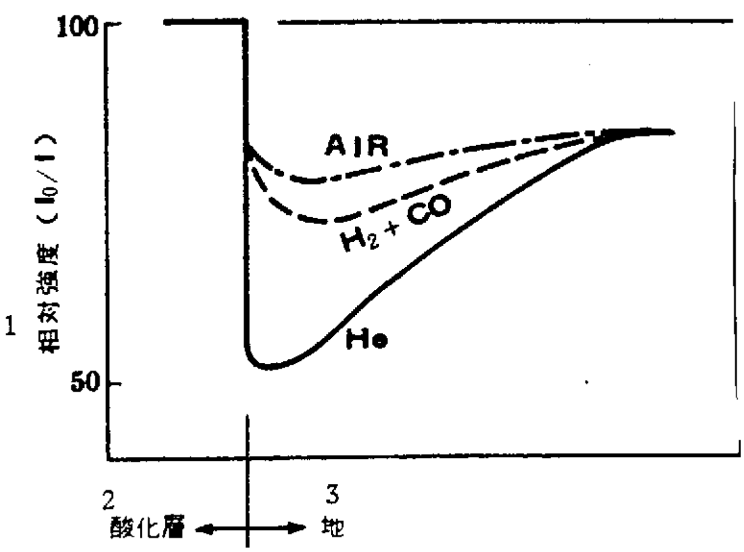

Fig. 2. Cr density in depth direction in Incoloy 800 corroded in various atmospheres in 1000°C.

Key: 1. Relative strength
2. Oxidized layer
3. Matrix

Fig. 3. Results of EPMA analysis of corroded layers of Incoloy 800 in various atmospheres at 1000°C (from Sakakibara et al.).

Key: 1. Corroded in air
2. Corroded in 65% CO + 35% H_2
3. Corroded in 99.99% He

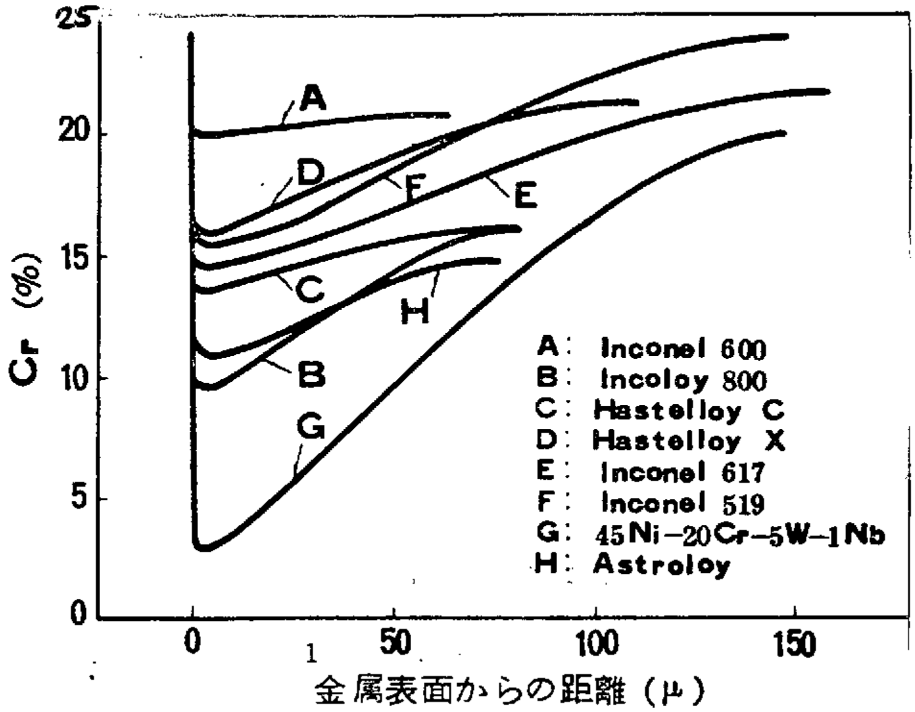

Fig. 4. Amounts of Cr on the metal surfaces of various alloys oxidized in He.

Key: 1. Distance from metal surface

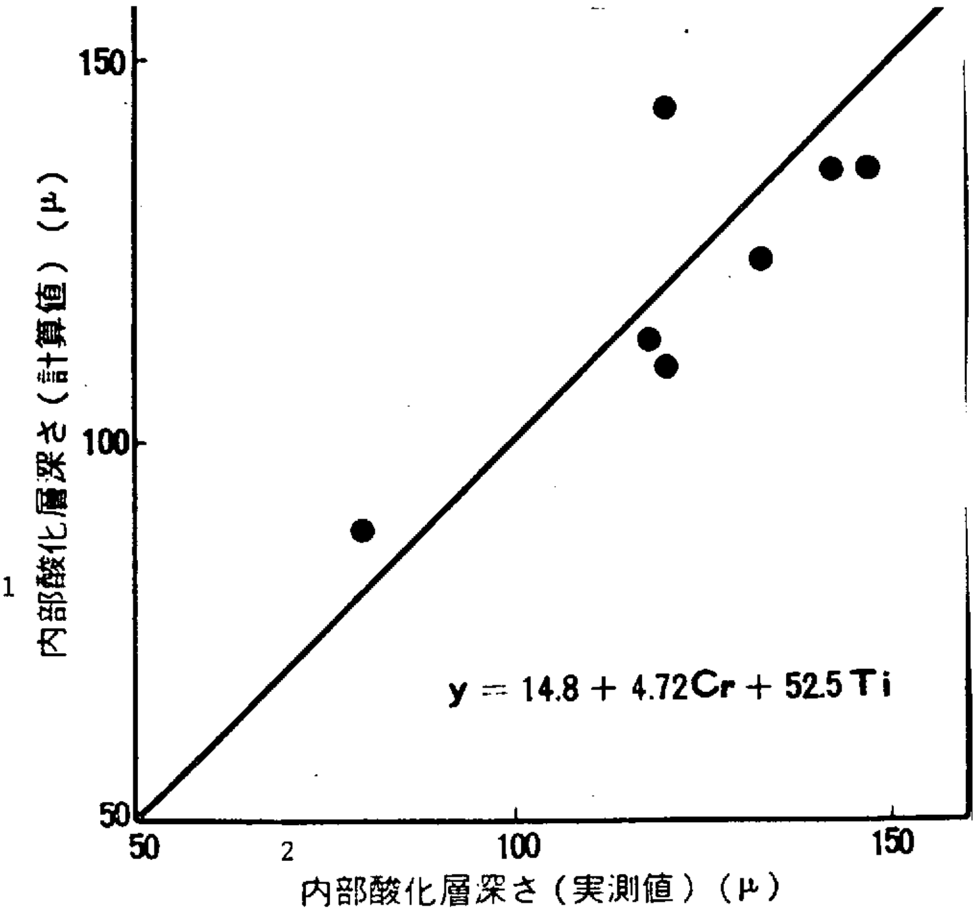

Fig. 5. Correspondence between measured and calculated values of depths of internal oxidized layers.

Key: 1. Depth of internal oxidized layer (calculated values)
2. Depth of internal oxidized layer (measured values)

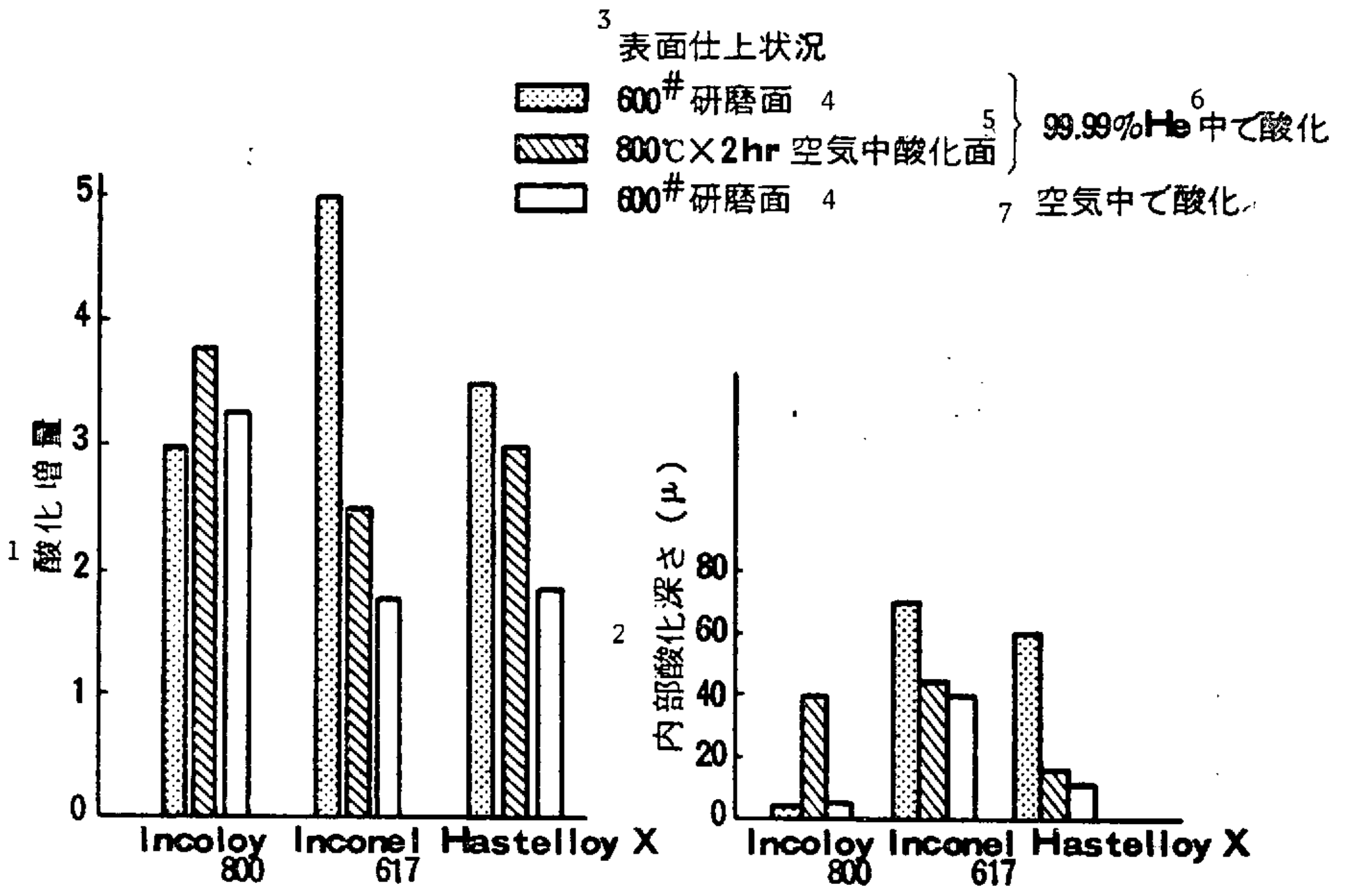

Fig. 6. Relationship between surface finishing of various alloys and oxidation weight increases and internal oxidation (oxidizing conditions: oxidized for 600 hours at 1000°C, in 99.99% He and in air).

Key:
1. Oxidation weight increases
2. Depth of internal oxidation
3. Surface finishing conditions
4. 600# polished surface
5. Surface oxidized in air at 800°C x 2 hours
6. Oxidized in 99.99% He
7. Oxidized in air

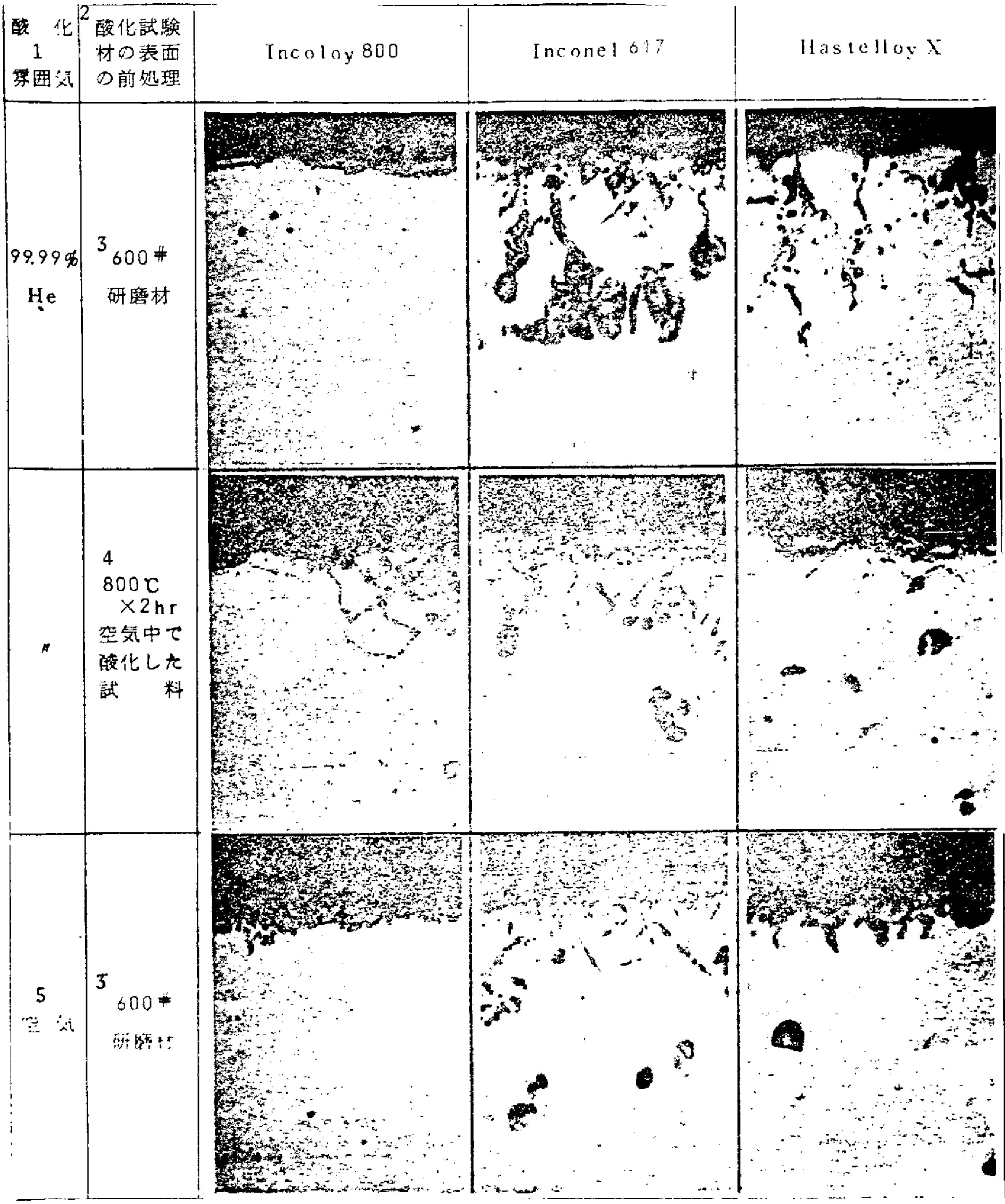

Fig. 7. Relationship between surface finishing of various alloys and internal oxidation (oxidized at 1000°C for 600 hours) (Magn. 500).

Key: 1. Oxidizing atmosphere; 2. Pretreatment of surface of oxidization test pieces; 3. 600# polished pieces; 4. Test pieces oxidized in air at 800°C x 2 hours; 5. Air.

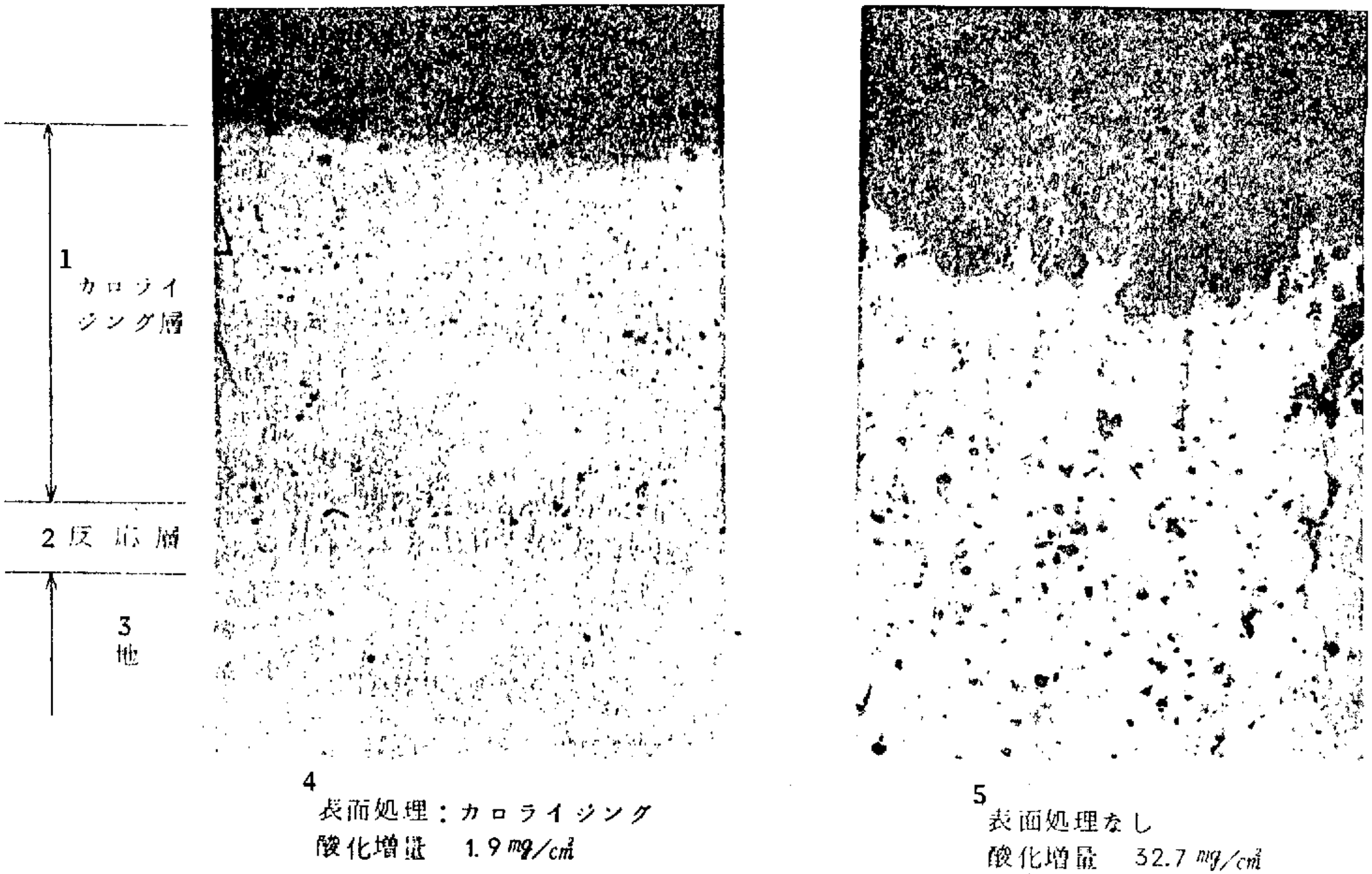

Fig. 8. Effects of surface treatment on oxidation in helium (600 hours in helium at 1000°C).

Key: 1. Calorizing layer
2. Reaction layer
3. Other
4. Surface treatment: calorizing
Oxidation weight increase:
5. No surface treatment
Oxidation weight increase:

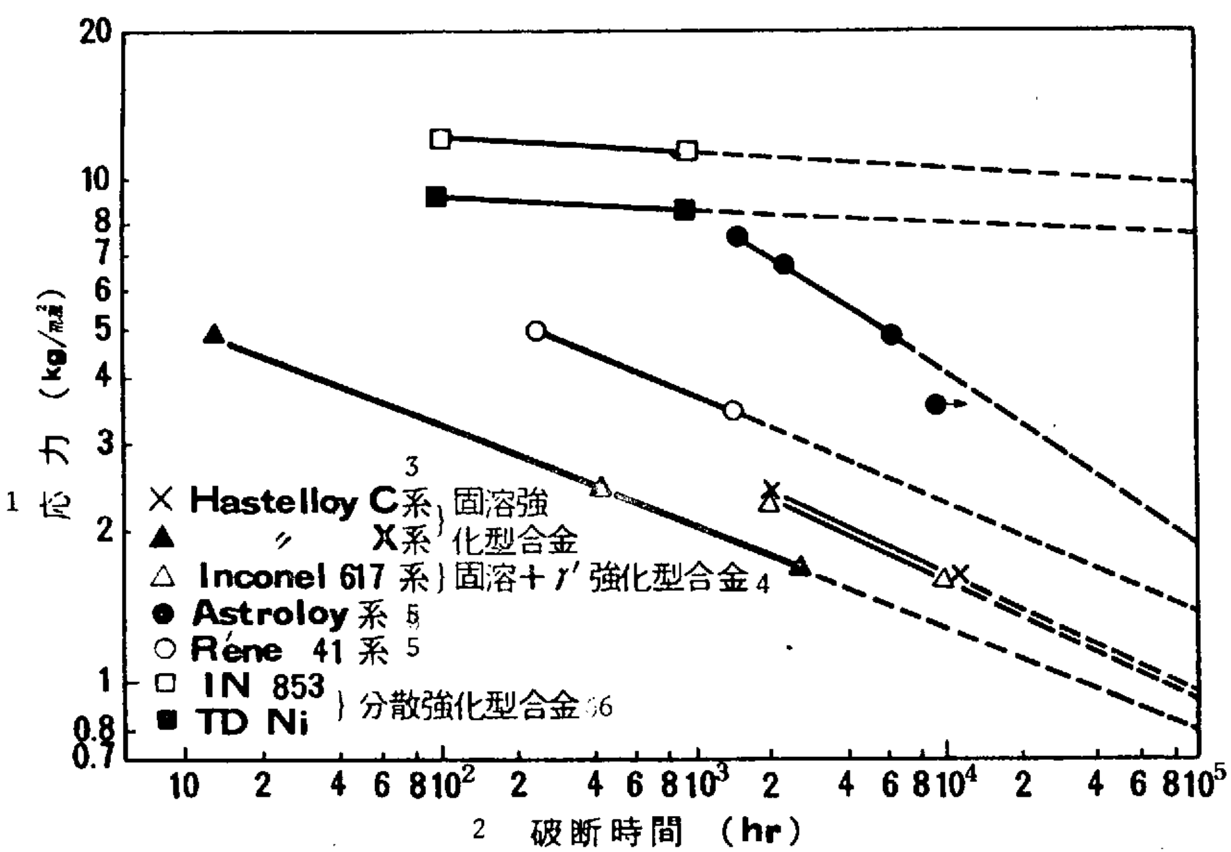

Fig. 9. Stress-rupture curves of Ni-base superalloys at 1000°C (from Sakakibara et al.).

Key: 1. Stress
2. Rupture time (hours)
3. Hastelloy C system: solid-solution-reinforced
" X system: type alloy
4. Inconel 617 system: solid-solution + γ'-reinforced type alloy
5. System
6. Dispersion-reinforced type alloy

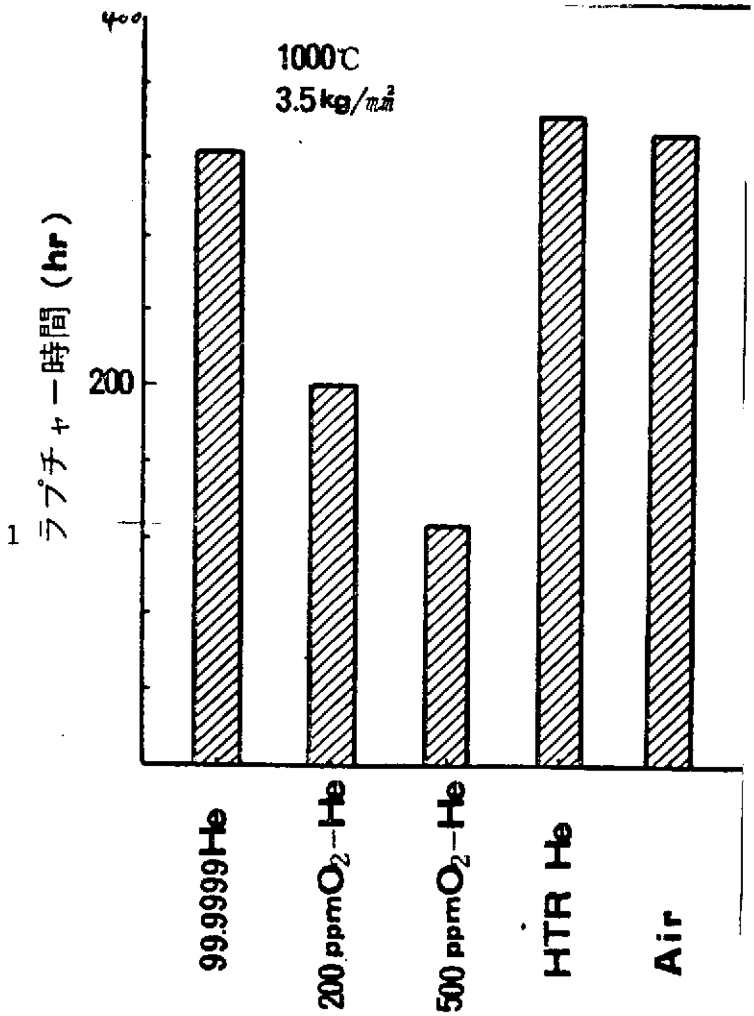

Fig. 10. Effect of impurity contents in He gases on the rupture times of Inconel 617.

Key: 1. Rupture time (hours)

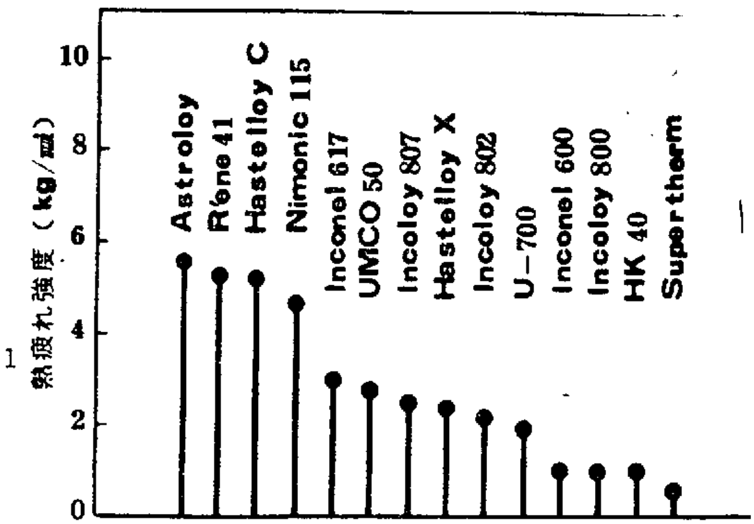

Fig. 11. Thermal fatigue strengths of various alloys (500 times between 200° and 1000°C).

Key: 1. Thermal fatigue strength

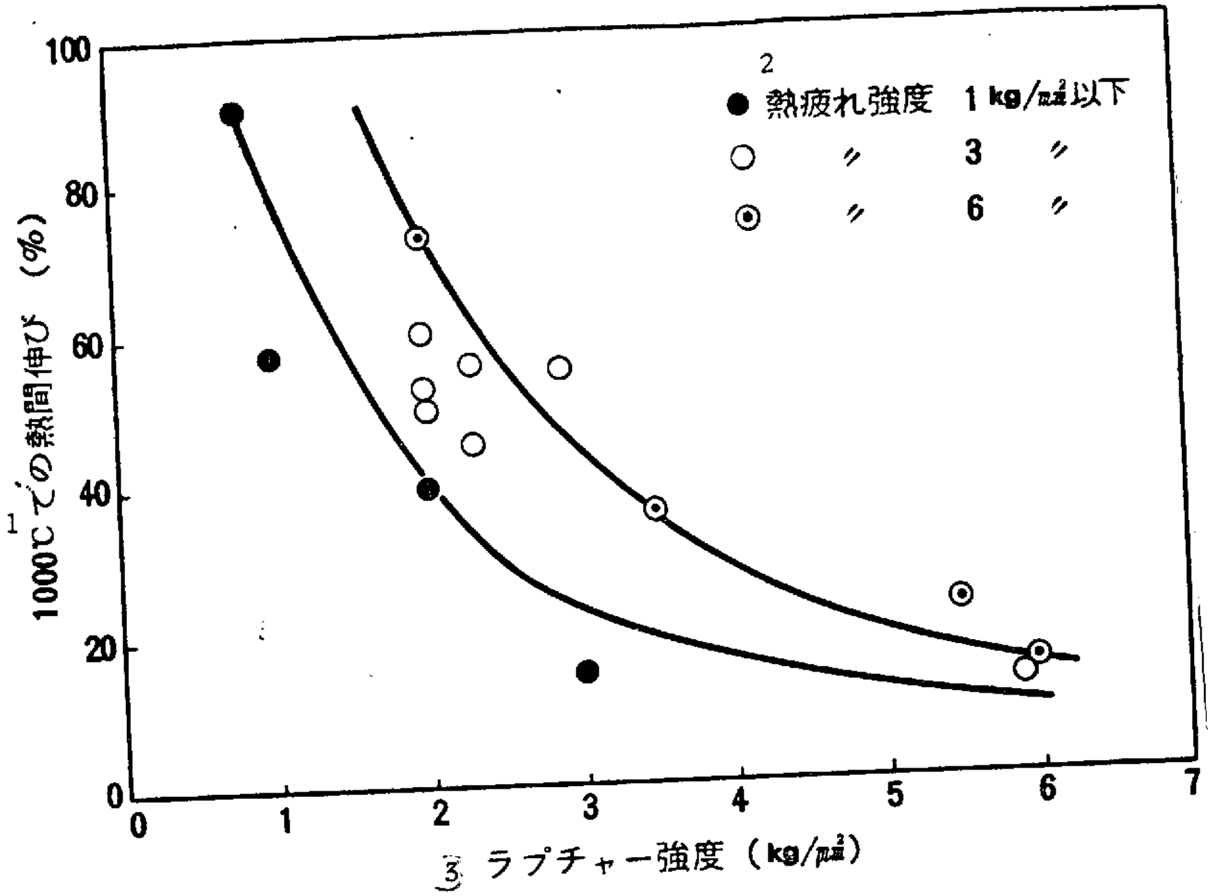

Fig. 12. Relationship between the rupture strength at 1000°C, 1000 hours, the hot elongation at 1000°C, and the thermal fatigue strength.

Key: 1. Hot elongation at 1000°C (%)
2. Thermal fatigue strength 1 kg/mm^2 or less
3. Rupture strength

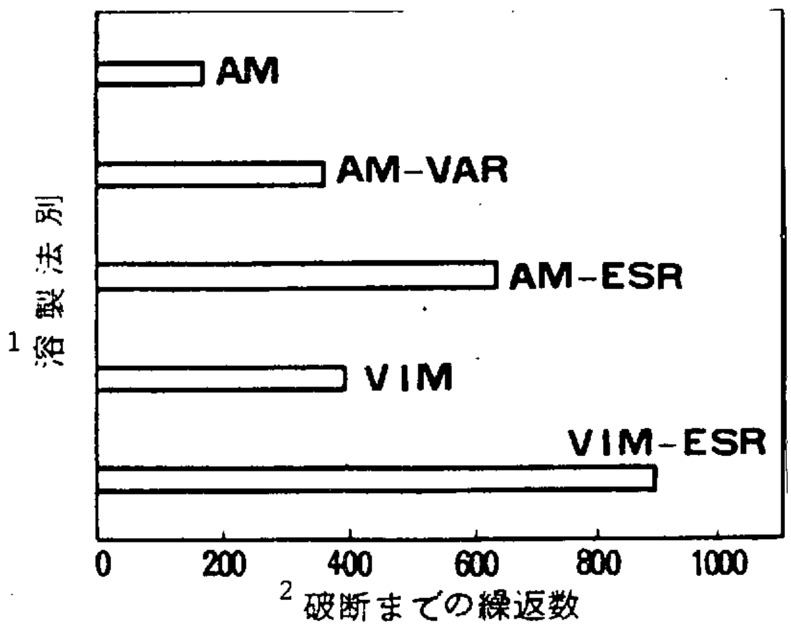

Fig. 13. Melting practices of Incoloy 800 and thermal fatigue rupture performance.

Key: 1. Melting practices
2. Number of repetitions until rupture

TABLE 2. EUTECTIC COMPOSITIONS AND EUTECTIC TEMPERATURES OF Ni BINARY ALLOYS.

Alloy System	Eutectic Compositions (wt %, at %)	Eutectic Tempera-tures (℃)
Ni−Aℓ	γNi, $Ni_3Aℓ$ (Ni 89, 79)	1385
Ni−O	Ni, NiO (O 0.24, 0.89)	1438
Ni−S	Ni, Ni_3S_2 (S 21.5, 33.4)	645
Ni−P	Ni, Ni_3P (P 11, 19)	880
	Ni_3P_2, Ni_2P (P 20, 32.2)	1106
Ni−B	Ni, Ni_2B (B 4, 18.5)	1140
	Ni_3B_2, NiB (B 13, 44)	990
Ni−Pb	γNi, Pb (Pb 34, 12.7)	1340
Ni−Zr	Zr, Zr_2Ni (Ni 17, 24)	961
Zr−Mn	Mn, $ZrMn_2$ (Zr 22.5, 32.5)	1135

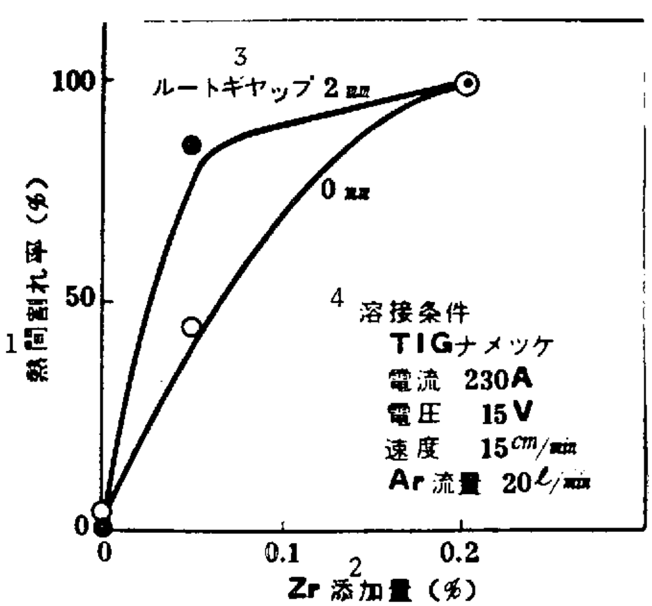

Fig. 14. Effects of addition of Ze upon high-temperature cracking in welding of Inconel 600 (circle batch tests).

Key:
1. Hot cracking ratio
2. Amounts of Zr added
3. Root gap
4. Welding conditions:
   TIG lick-on
   Current
   Voltage
   Velocity
   Air flow rate

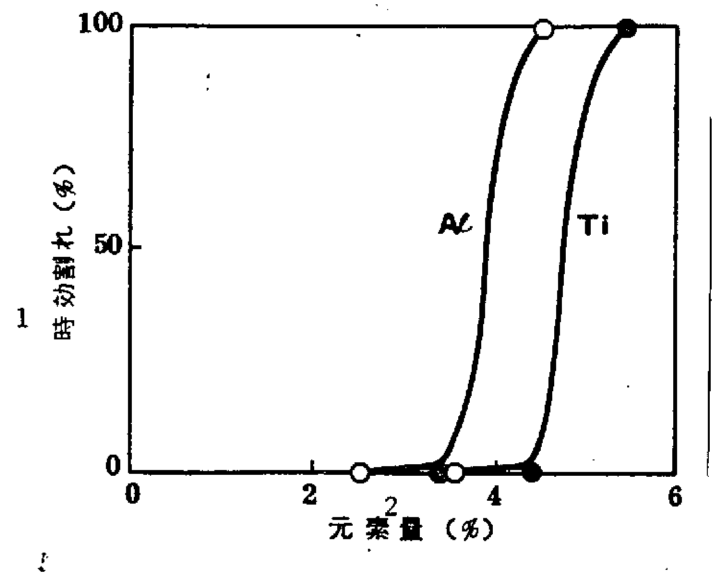

Fig. 15. Relationship between amounts of Al and Ti added to Astroloy alloys and aging cracking in welding (welding conditions: TIG lick-on, 230 A, 15 V, 15 cm/min, 20 l/min Ar) (from Sakakibara et al.).

Key: 1. Aging cracking
2. Amounts of elements

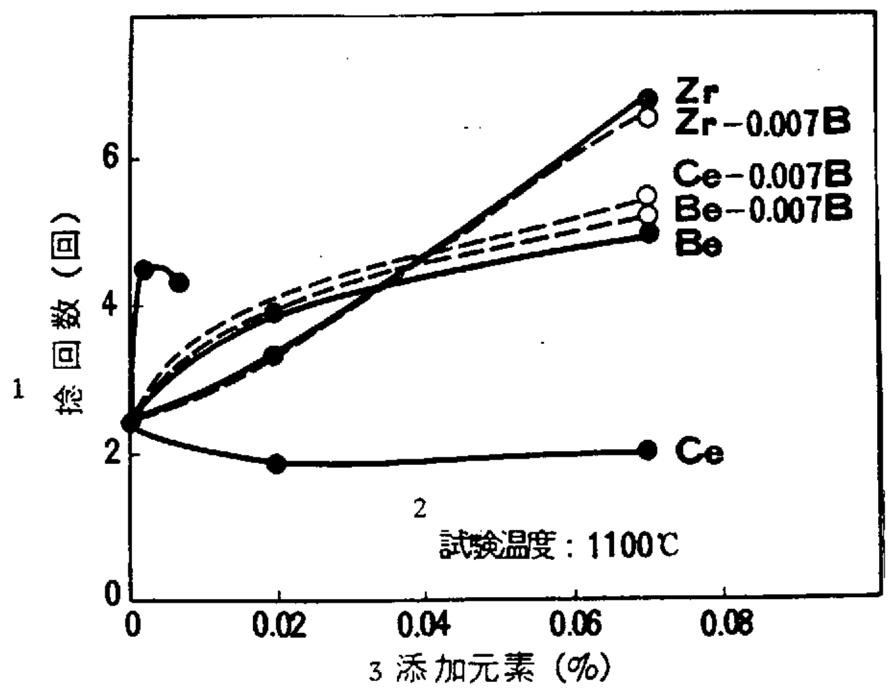

Fig. 16. Relationship between hot torsion numbers in Inconel 600 and microelements added.

Key: 1. Torsion number (number of times)
2. Test temperature
3. Added element

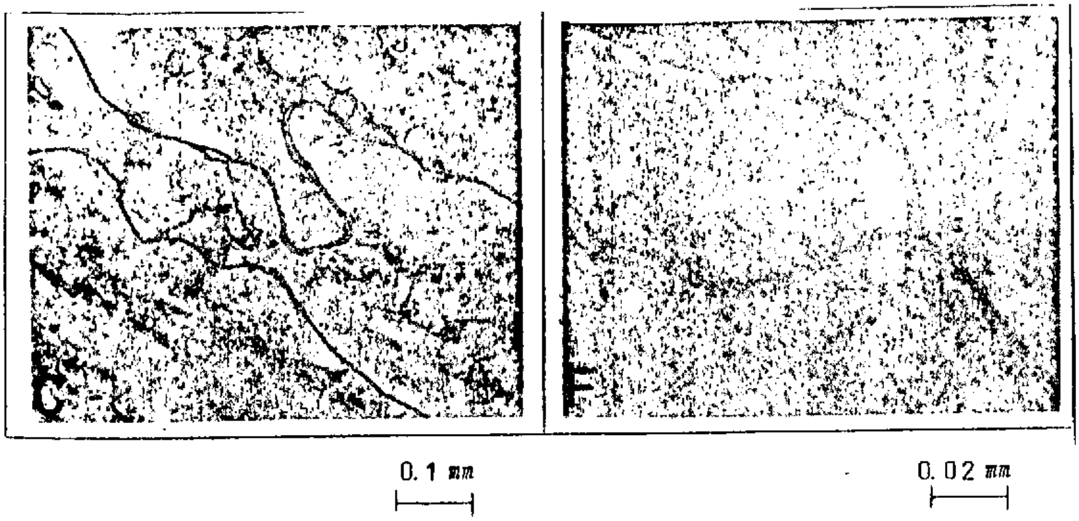

Fig. 17. Fission track photographs of Inconel 600 to which B-Zr was added.

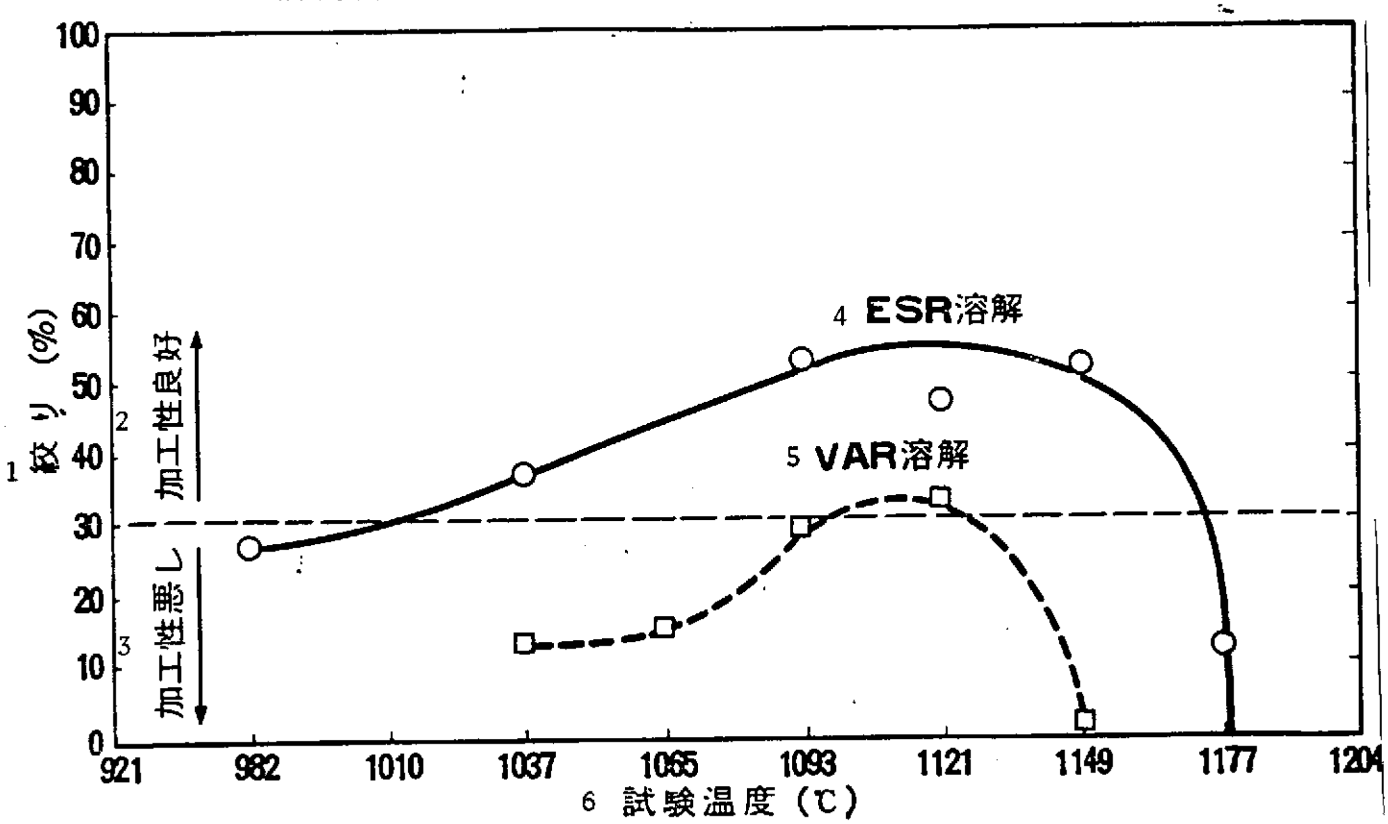

Fig. 18. Influence of melting practices on hot ductility in Udimet 700 by Greeble tests (from J.W. Pridgeon et al.).

Key: 1. Reduction of sectional area
2. Good workability
3. Poor workability
4. ESR melting
5. VAR melting
6. Test temperature

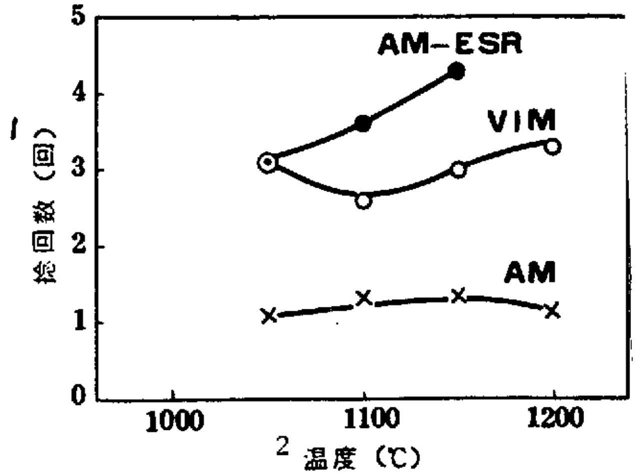

Fig. 19. Melting practices of Hastelloy X and hot torsion numbers.

Key: 1. Torsion numbers (number of times)
2. Temperature

TABLE 3. COMPARISON OF MELTING AND REFINING PROCESSES IN HEAT-RESISTANT ALLOYS (R. SCHLATTER).

	1 Primary Melting Process					2 Secondary Melting Process			
	AM	AM+VR	AM+AOD	VIM	PAM	VAR	EFR	EBM	PAR
Versatility	-	+	+	++	++	++	+	+	++
Alloying	++	++	++	++	++	-	-	-	-
Super-high temperature properties	+	+	+	++	++	-	-	++	++
Reaction with refractories	-	-	--	-	(--)	++	++	++	++
Slag processing	+	+	++	-	+	--	++	--	+
Regulation of Composition	+	+	+	++	++	++	+	+	+
Degassing properties	--	+	+	++	(+)	++	--	++	+
Deoxidation, decarbonization	-	++	+	++	(+)	+	-	++	(+)
Desulfurization	++	+	++	-	+	-	++	-	+
Decarbonization	-	++	++	++	(+)	+	-	++	(+)
Evaporation of impurities	--	+	+	++	(-)	+	--	++	(+)
Degree of cleanness	--	+	+	++	++	++	+	++	++
Control of solidified structures	--	--	--	--	--	+	+	++	++

Evaluation: ++ Superior - Poor ( ) Estimated effect
+ Good -- Indadmissible

TABLE 4. SUITABILITY CRITERION OF VARIOUS SUPERALLOYS FOR HIGH-TEMPERATURE GAS REACTOR MATERIALS.

Characteristics / Alloys	-1000℃·10^5hr Estimated Rupture Strength	1000℃·600hr 99.99% He Oxidation Wt. Increase	200~1000℃ Rupture Strength in Heat Cycles	Weld Cracking Score	Hot Torsion Rupture No. (No. of Rotations	Total Score	Overall Judgement
Incoloy 800	6	1	6	2 (6)	2 (4)	23	B
〃 807	5	4	4	2 (6)	5 (10)	29	C
Inconel 600	7	2	6	1 (3)	2 (4)	22	B
〃 617	4	4	3	1 (3)	3 (6)	20	B
〃 625	5	1	4	3 (9)	6 (12)	31	D
Hastelloy C	4	2	2	3 (9)	4 (8)	25	C
〃 X	4	3	4	2 (6)	3 (6)	23	B
René 41	3	4	2	5 (15)	6 (12)	36	E
Nimonic 115	2	3	3	5 (15)	6 (12)	35	E
Astroloy	2	1	4	5 (15)	4 (8)	30	D
HK 40	4	5	5	4 (12)	5 (10)	36	E
Supertherm	3	5	5	4 (12)	6 (12)	37	E
TD－Ni	1						
Inconel 853	1						

10 ※1 評点は非常に優れているものを1とし，非常に悪いものを7としてつけた。

11 ※2 総合判定はA：優，B：良，C：可，D：困難，E：非常に困難

12 ※3 （ ）内は荷重をかけた数字。

1) Scoring. Extremely good: 1; extremely bad: 7.
2) Overall judgement. A: excellent; B: good; C: passable; D: difficult; E: extremely difficult.
3) Numbers enclosed in parentheses represent the weights assigned.

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